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DEVELOPMENT OF A MODULATED-MICROSTRUCTURE HEAT
TREATABLE STEEL

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Second

Annual Technical Report

on

DEVELOPMENT OF A MODULATED-MICROSTRUCTURE
HEAT TREATABLE STEEL

by

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July 10, 1975

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I. INTRODUCTION

The objective of this research is to design and develop a modulated-microstructure steel consisting of alternating hardened layers and soft layers such that the complex has a tensile strength of 400 ksi (2.8 KN/mm^2) and fracture toughness of 150 ksi$\sqrt{in}$ (160 N/mm$^2\sqrt{m}$). The alloy complex presently under development is to combine layers of high-speed steel with layers of partially austenitic steel. The full strength of the hard layer is to be usable because the ductility of the softer layer is to provide toughness which inhibits crack propogation. This alloy is to be capable of being softened for machining and hardened by uniform heat treatment. Its development represents a very attractive goal in view of the presently available properties of alloy steels which have been summarized by Lange[1] and presented in Fig. 1. His estimate of the high strength technological limit of fracture toughness is substantially less than 100 ksi$\sqrt{in}$ (.11 KN/mm$^2\sqrt{m}$) at stress levels in excess of 300 ksi (2.1 KN/mm^2). Thus our design objective represents a very desirable target and progress is being made towards its achievement.

Reported in the First Annual Technical Report[2] (1973-1974) was the development of a technique (hot rolling) to fabricate the alloy complexes. A design rationale was formulated and applied to a model system whose hard layer is an 01 commercial tool steel and whose soft layer is a specially designed alloy (PS2). This system was manufactured and tested as reported in that report. An important step in the fabrication process was to insure that the alloy layers will partially debond when they deformed heavily or fractured and hence improve the fracture toughness of the complex. In the model system, partial debonding was accomplished by purposely introducing oxide particles into the interface. In view of the importance of having
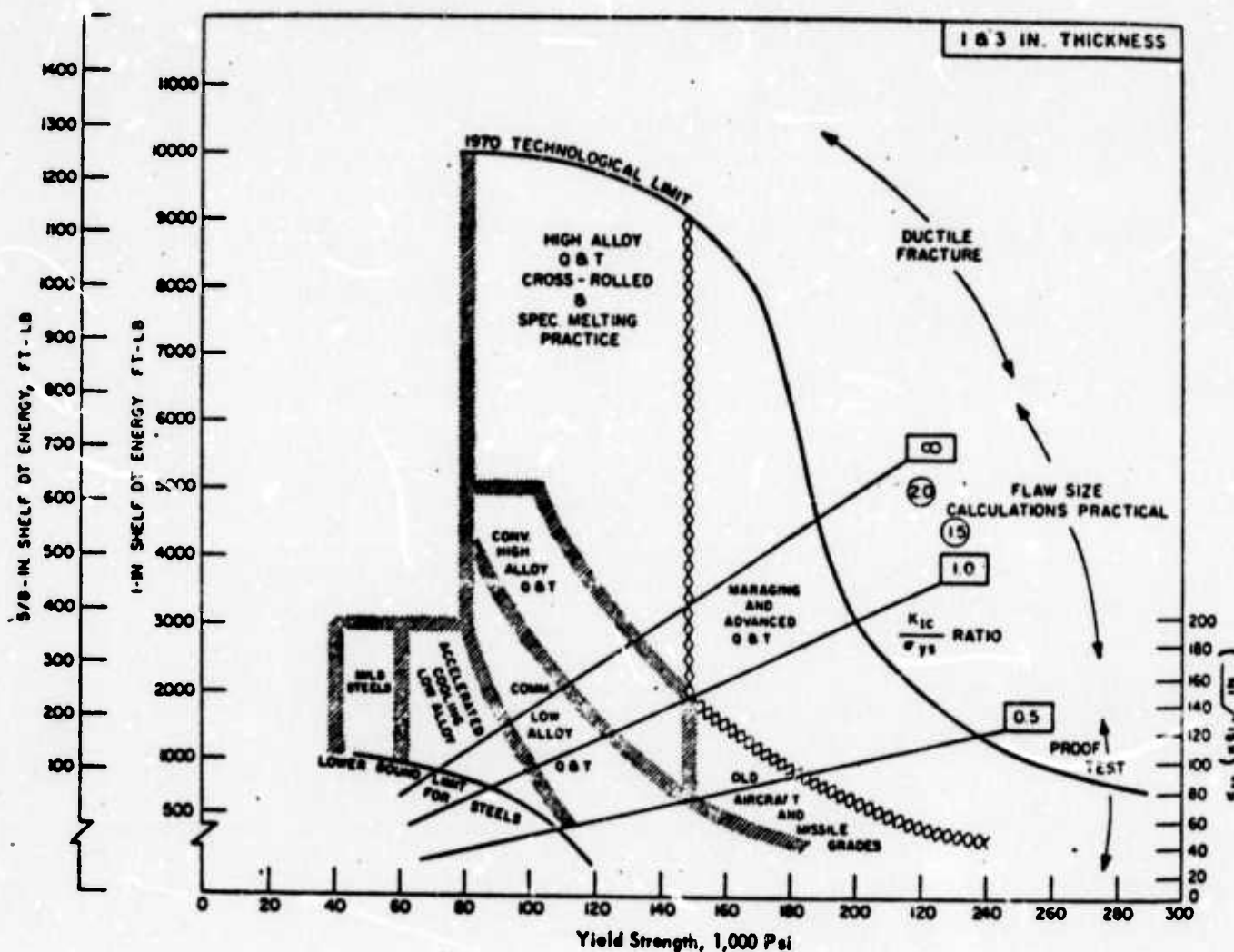
I & 3 IN. THICKNESS

1970 TECHNOLOGICAL LIMIT

HIGH ALLOY
Q & T
CROSS - ROLLED
&
SPEC. MELTING
PRACTICE

DUCTILE
FRACTURE

CONV.
HIGH
ALLOY
Q & T

MARAGING
AND
ADVANCED
Q & T

∞

20

15

FLAW SIZE
CALCULATIONS PRACTICAL

10

$\frac{K_{IC}}{\sigma_{ys}}$ RATIO

MILD
STEELS

ACCELERATED
COOLING
LOW ALLOY

CONV.
LOW
ALLOY
Q & T

0.5

LOWER BOUND LIMIT FOR STEELS

OLD
AIRCRAFT
AND
MISSILE
GRADES

PROOF
TEST

S/8-IN SHELF DT ENERGY, FT-LB

1-IN SHELF DT ENERGY, FT-LB

K_{IC} (KSI $\sqrt{in}$)

Yield Strength, 1,000 Psi

Fig. 1. Summary of Mechanical Properties of Alloy Steel and the 1970 Technological Limit, After Lange (1971). The target strength of 400 ksi and toughness of 150 ksi√in. are well above this limit.

the layers to decouple in order to retard crack propogation and thus improve toughness, further study of the interfacial bonding was carried out in the second project year. Guided and encouraged by the relatively high strength and good toughness obtained in the O1/PS2 model system, a high strength system incorporating the improved interface was designed. This report describes the progress in the development of this system.

The development of the alloy complex has proceeded through: the selection of the component alloys and their production (Alloy Selection), the development of a suitable interface (Interface Design), the determination of an approximately optimum heat treatment (Heat Treatment), and the initial testing and evaluation of the resulting complexes (Results and Discussion). The last stage and the expected feedback is presently in process.

II. ALLOY SELECTION

The success of the new alloy complex depends on the selection of a suitable pair of alloys and a suitable interface. Briefly they must be (1) compatible in heat treatment, that is they must be subject to a common heat treatment for hardening and tempering; (2) they must be equal in carbon activity because all reasonable fabrication processes involve time for substantial interlayer carbon diffusion; (3) they must be bonded at the interface with a phase which is easily cracked so that the soft layers can shear without mechanical constraint from the adjoining hard layers.

II.1 Hard Layer

In order to design a system of higher strength than the O1/PS2 model system, it was necessary to select as the hard layer a steel of maximum strength. A commercial high speed tool steel REX 71 was chosen. In its triple tempered condition, it has a hardness of R_c 69.6 (DPH 1015). This high hardness implies a tensile strength of about 520 ksi (3.6 KN/mm^2) on the basis of the experimental relationship between yield strength and hardness reported for tool steel[3]. The composition of this steel is listed in Table I.

II.2 Soft Layer

Theoretical analysis[2] indicates that the toughness increases with f_s, the fraction of soft layer. As f_s increases, the strength of the soft layer must also be increased to maintain a given strength of the complex, and as the strength of the soft layer increases, its ductility decreases under plane strain conditions. It was shown[4] that maximum toughness for several steels occurs at a strength of 150 to 200 ksi(1 KN/mm^2 to 1.4 KN/mm^2). With this in mind, and with the REX 71 tool steel already selected to be the

TABLE I

Composition of Crucible CPM REX 71 High Speed Tool Steel

C	Mn	P	S	Si	Cr	V	W	Mo	Co
1.17	.32	.009	.014	.29	3.63	1.27	10.06	5.26	12.10

TABLE II

Composition of PS4 Alloy

C	Co	Ni	V	Mo
.28	10	26	.5	1.5

hard layer, a special alloy PS4 was designed. It was designed to have a M_s temperature about 0°C so that it can be presented with unstable austenite and can be subzero quenched to produce a mixture of austenite and martensite with a strength of 150 ksi (1 KN/mm^2). Use was made of the known carbon activity coefficienty[5] in the design so that the soft and hard layer have the same carbon activity coeficients. The composition of this alloy is shown in Table II.

III. INTERFACE DESIGN

Debonding at the interface played an important part in the success achieved in the model system[2]. Heavy emphasis is therefore given to designing the interface in the high strength system. In the model system, partial debonding between layers was accomplished by the introduction of oxide particles into the interface. The oxide particles were introduced by controlled oxidization of the O1 surface, by oxidization in air at 1110°F (600°C) for 2 hrs. The hard/soft layer interface was bonded by repeated hot rolling. The complexes so produced had rolling reduction of about 60%. Two variables influencing interface bonding were investigated. Rolling reduction and post rolling anneal were studied experimentally on the model system. In order to achieve the optimum desired bonding between the surfaces, a series of experiments were performed to measure the bonding energy (the energy required to separate a unit area of the surfaces) of O1/O1 surfaces. Samples containing such surfaces were produced with rolling reductions varying from about 4% to 50%. The experimental method used was essentially that described previously[2]. These experiments produced the rather surprising result that the surface bonding energy is essentially independent of the amount of rolling reduction and is constant at about .05 J/cm^2. See Fig. 2. It was postulated[6] that the surface bonding energy decreases only slowly with increased amount of oxides at the interface and does not vary significantly with increased rolling. Rolling may well influence the interlayer bond strength by reducing the oxide particle size and hence reducing the size of the longest initial crack. Further experiments showed that the surface bonding energy depends on the post rolling heat treatment as shown in Fig. 3.

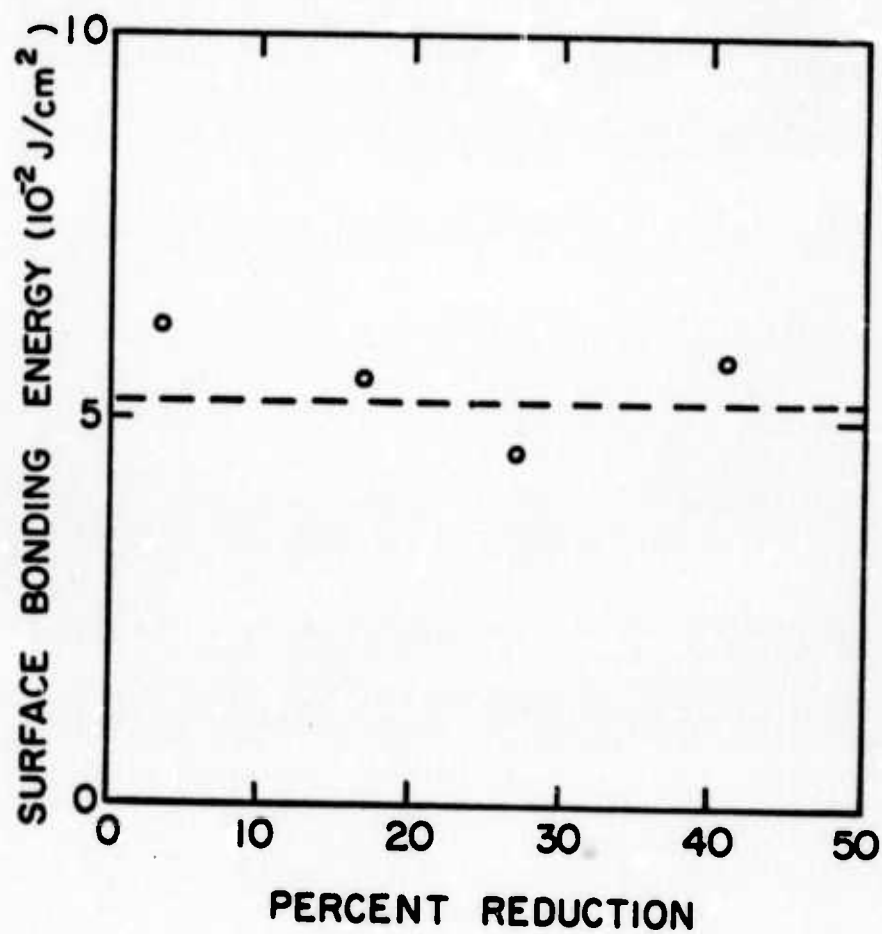A value of .05 J/cm^2 for γ, the surface bonding energy, for the oxidized

Fig. 2: Bonding of Oxidized 01/01 Interfaces vs. Reduction by Hot Rolling. Following rolling all samples were annealed for ten minutes at 2100°F (1150°C).
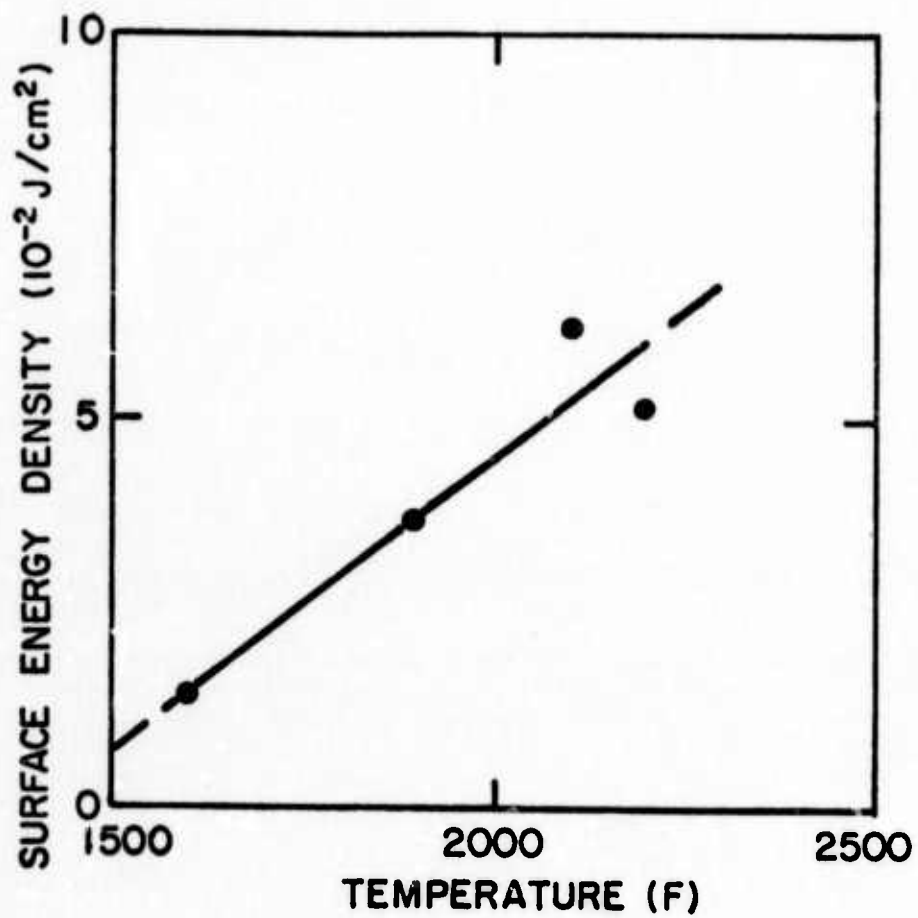
Fig. 3: Bonding of Oxidized 01/01 Interfaces vs. Heat Treatment Temperature
(Ten Minute Anneals) Following Rolling. The rolling reduction is
5 ± 2%.

01/01 interface studies above implies a value of stress intensity factor of about 15 ksi$\sqrt{in}$ (16 N/mm$^2\sqrt{m}$) for the propogation of a crack along the interface. A significant improvement in tensile ductility was achieved in the model system when this kind of interface was substituted for well bonded ones, however, for the high strength system, further reduction in the value of γ was desired. A 30% reduction in the value of γ for the 01/01 interface is obtained by inserting a 0.002" thick stainless steel foil which has been oxidized in air at 1740°F (950°C) for 15 minutes. Unfortunately, such an interface design was not suitable for the high strength system because after annealing large bubbles are present at the interface. A possible explanation is that after the interface (R71/SS/R71) has been produced following hot rolling, the oxides react with the carbon dissolved in REX 71 during heat treatment at 2100°F (1150°C). This reaction produces gases (CO) which develop enough pressure to rupture the interface. In an attempt to overcome this difficulty for the interface between REX 71 and PS4 layers, only the side of the stainless steel foil which faces the PS4 layer was oxidized. The interface so produced by hot rolling is too well bonded, a value of γ of 0.1 J/cm^2, which means that the bonding between the stainless steel and PS4 is stronger than the corresponding bonding between stainless steel and 01 steel. To reduce the value of γ, two layers of stainless steel foils were used. Each foil has only one surface oxidized and these surfaces face each other while the unoxidized surfaces face the REX 71 or the PS4 surfaces. After rolling, the REX 71 and PS4 form well bonded interfaces. See Fig. 4. Indeed, interfaces so produced give interface bonding energy between .015 and .025 J/cm^2, indicating that this interface design has a bonding energy which is less than half of that of the model system. This interface design was adopted for the high strength system.

Fig. 4: Trial Interface REX 71/Stainless Steel/REX 71. The
        stainless steel was oxidized only on the side facing
        the other stainless steel. Surface bonding .013 J/cm^2

2% Nital Etch   1280X

The stainless steel foils are roll bonded to both the hard layer and soft layer materials as these materials are rolled down to sizes of about 0.020" thick from initial stock sizes of about 3/8" thick. The initial stock was wrapped in several layers of 0.002" thick stainless steel foils so that the stainless steel makes up about 5% of the whole volume. The stainless steel foil wrappings are then heli-arc welded and the entire assembly is sandwiched between 1/8" thick stainless steel plates and hot rolled. The result is that the REX 71 or PS4 sheets so produced have a layer of stainless steel bonded to them. The inside surface of the stainless steel is unoxidized and is well bonded to the REX 71 or PS4. The outside surface of the stainless steel wrapping oxidized during hot rolling and experience indicates that the amount of oxidization is satisfactory. The REX 71 interface exhibits an affected zone which could be due to decarburization by the stainless steel. When the stainless steel foils were precarburized in a controlled atmosphere of methane and hydrogen to attain a carbon potential equal to that of the REX 71 (or PS4), the affected zone was substantially reduced in thickness.

IV. HEAT TREATMENT

The REX 71 steel is obtained in bars 3/8" x 1" (9.5 mm x 25 mm). For the manufacture of MMS complexes, the bars have to be rolled down to sheets of .020" (0.5 mm) thick at a temperature of about 2100°F (1150°C). The PS4 alloy is cast in ingots of about 1 1/4" x 3" x 10". The ingot is cut and rolled down to sizes in a similar manner. According to manufacturer's instructions, the REX 71 steel must be annealed after rolling by holding at 1600°F (870°C) for 2 hrs. and cooling slowly at 25°F (14°C) per hour in the furnace to below 1000°F (540°C). This process is performed on all complexes subject to mechanical testing. Each time REX 71 is hot rolled, either in the original 3/8" x 1" bars or in the form of stacks of sheets to make the complexes, this elaborate heat treatment is always performed. The slow cooling rate was controlled by a Model FGE 5500 Data-Trak programmer and the usual furnace controller.

Hardening heat treatment of the complex is essentially the heat treatment for the REX 71 recommended by the manufacturer. It requires austenitizing at 2175°F (1190°C) for 2 minutes, quenching in oil and tempering at 1000°F (540°C) for two hours, cooling to room temperature and repeating this tempering treatment two times. This is the so-called triple tempering at 1000°F. The PS4 alloy is designed to have compatible heat treatment and it can be presented in a number of different micro-structures for testing and all these structures are attainable by heat treatments compatible with the maximum hardness in the REX 71 hard layers. A diagram illustrating some of the possible heat treatments is shown in Fig. 5. The sequence ABCE is required to harden REX 71, the insertion of subcool D is optional and subcool F and light temper G are without substantial effect on REX 71. The PS4 alloy is austenitized in Step A and remains austenitic until subzero cooling.
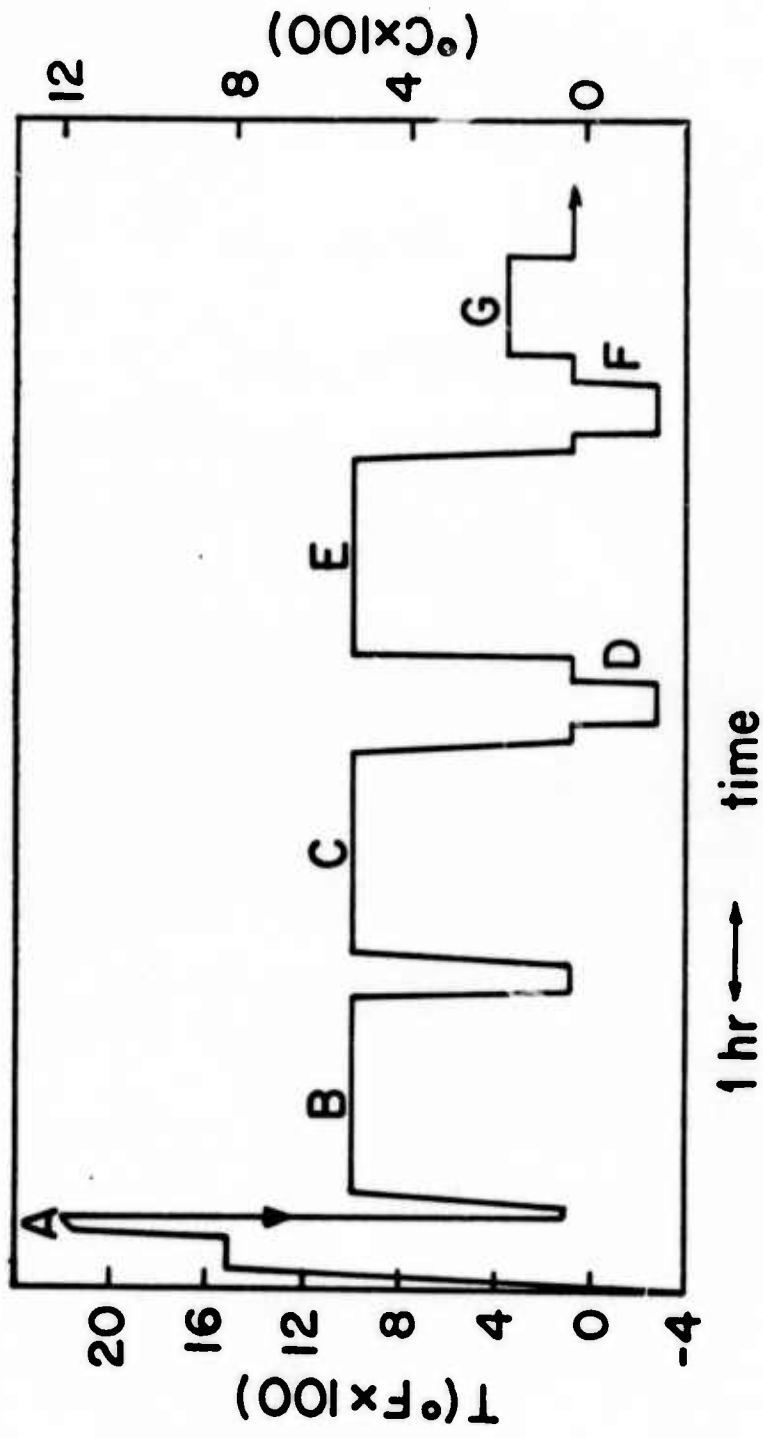
Fig. 5: Schematic Heat Treatment Schedule for the High Strength System.

V. RESULTS AND DISCUSSION

V.1. Properties of the Soft Layer Alloy, PS4

The important properties of PS4 are its heat-treatment compatibility with a high-speed steel such as REX 71 and its strength and ductility. The properties examined are: carbon activity, M_s temperature, response of austenite to tempering up to 1000°F (540°C), and response of martensite to tempering up to 1000°F (540°C). Tempering of austenite was followed by metallographic examination for carbide precipitation and by the variation of M_s. Tempering of martensite was followed by changes in hardness and fracture behavior.

A small quantity (1 kg) of PS4 alloy was melted at Purdue University's crystal growing facility. The procedure was described in the First Annual Technical Report. The desired characteristics, e.g., martensite start temperature, M_s, at or near room temperature and carbon activity matched to REX 71 were verified and a larger quantity (14 kg or 30 lbs.) were vacuum melted by the Armco Steel Co. and supplied to the project. Several determinations of the M_s temperatures of this PS4 yield M_s = 12°C.

Recently published[5] carbon activity coefficients have been used in the design of PS4 to match its carbon activity with that of the REX 71 hard layer. To check whether or not the carbon activity of PS4 is indeed equal to that of REX 71, several REX 71/PS4 complexes were prepared by hot-rolling. They were then enclosed in evacuated quartz capsules and annealed for 20 hrs. at a temperature of 2100°F (1150°C) and quenched. This time and temperature allows carbon diffusion but little diffusion of substantial solutes. The M_s temperatures of PS4 in these samples were then determined. They were found to be slightly above room temperature. This indicates that appreciable carbon had not diffused into or out of the PS4 and hence the carbon activities in the two alloys are nearly equal.

Several samples of PS4 alloy were given the ABCE heat treatment of Fig. 5. During this repeated tempering, none of them showed significant carbide precipitation in the PS4 austenite as evidenced by the fact that the PS4 M_s remained at about room temperature. The hardness of the PS4 alloys thus treated is about DPH 190. However, if the PS4 is liquid nitrogen quenched after the austenitization A and then tempered as in BCE its hardness is increased to DPH 450. Moreover other hardnesses are also attainable.

A set of as-cast PS4 samples was given the ABCEFG heat treatment of the schedule shown in Fig. 5. In this set the final subzero quench temperture was -321°F (-196°C) resulting in 78% martensite and the final tempering temperature, T_G, was varied from 100°F (40°C) to 1100°F (590°C). Fig. 6 shows the resulting hardness. The as quenched (F) hardness is 410 DPH. Tempering G results in softening to 360 DPH followed by hardening to 510 DPH. This secondary hardening occurs above 600°F (315°C). According to qualitative observations of fracture of notched specimens, secondary hardening in PS4 is accompanied by significant embrittlement. Fractographs show that the dimple fracture at low hardness, Fig. 7, is replaced by parting along prior austenite grain boundaries, Fig. 8, at peak hardness. Consequently in the complexes, tempering of martensitic PS4 will be restricted to temperatures below 500°F (260°C).

Two tensile tests were conducted on solid PS4 tensile bars rolled and heat treated in all essential respects as would be required in the manufacture of the complexes. The final heat treatment, ABCE of Fig. 5 left the PS4 nearly 100% austenite. Standard tensile bars, Fig. 9, were pin loaded by the previously described[2] grips on an 10,000 lb. Instron tensile machine. Strain was measured with an extensometer for small strains or calculated from the cross-head motion (0.02 in/min or .51 mm/min) for large strains. The resulting engineering stress-strain curves are shown in Fig. 10. Their
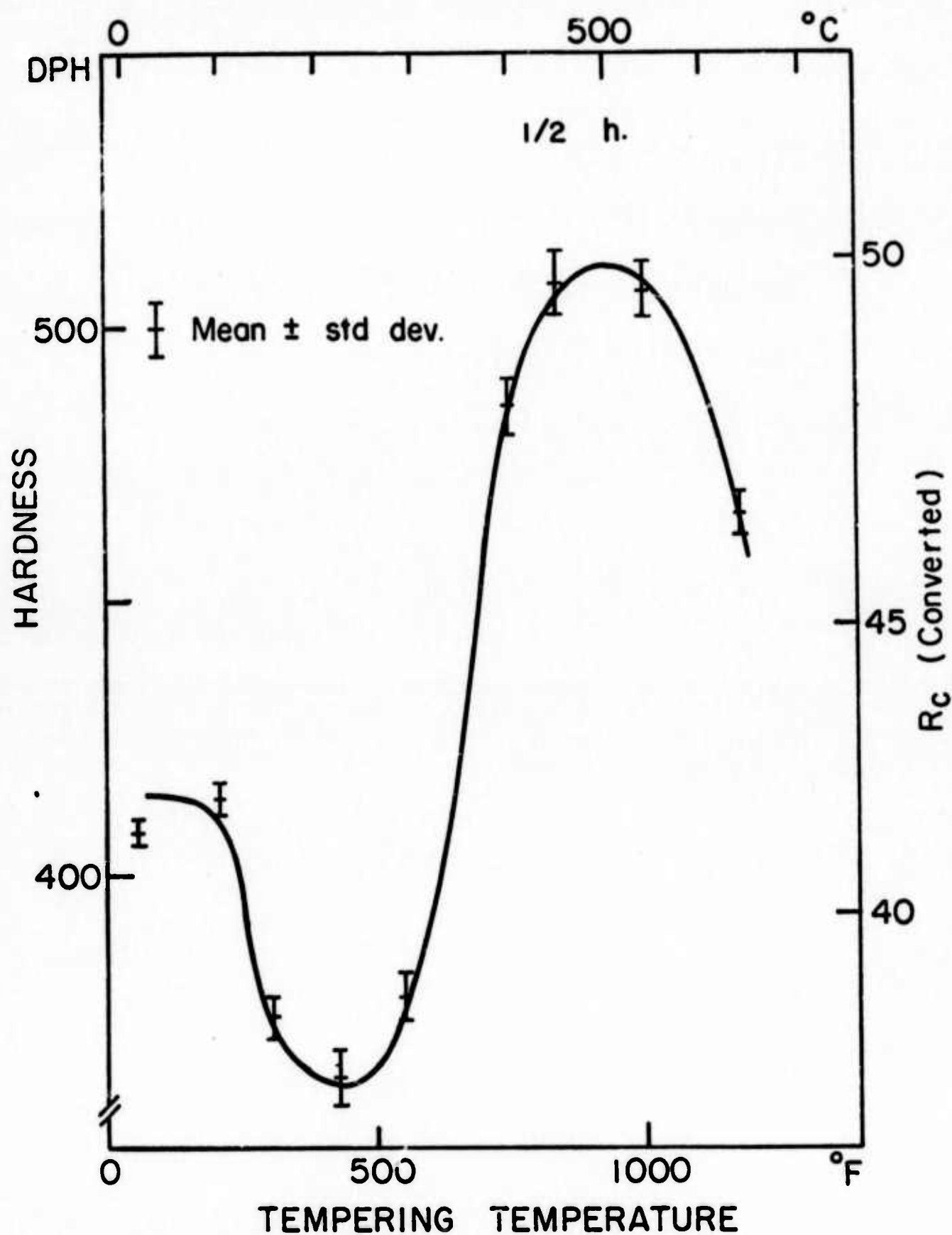
Fig. 6: Hardness of Tempered PS4 Alloy. Cast alloy hardened by austenitizing, at 2175°F quenched, and reheating three times at 1000°F and then cooling in liquid nitrogen to form martensite.

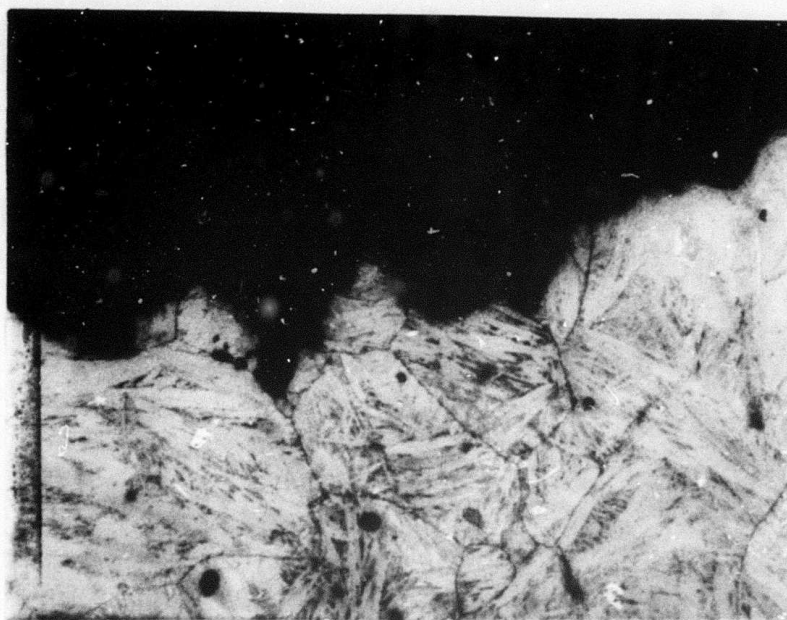Fig. 7A: Metallographic Section Through Impact Fraction of PS4
Tempered at 300°F (150°C) for 1/2 Hour.

2% Nital Etch   256X

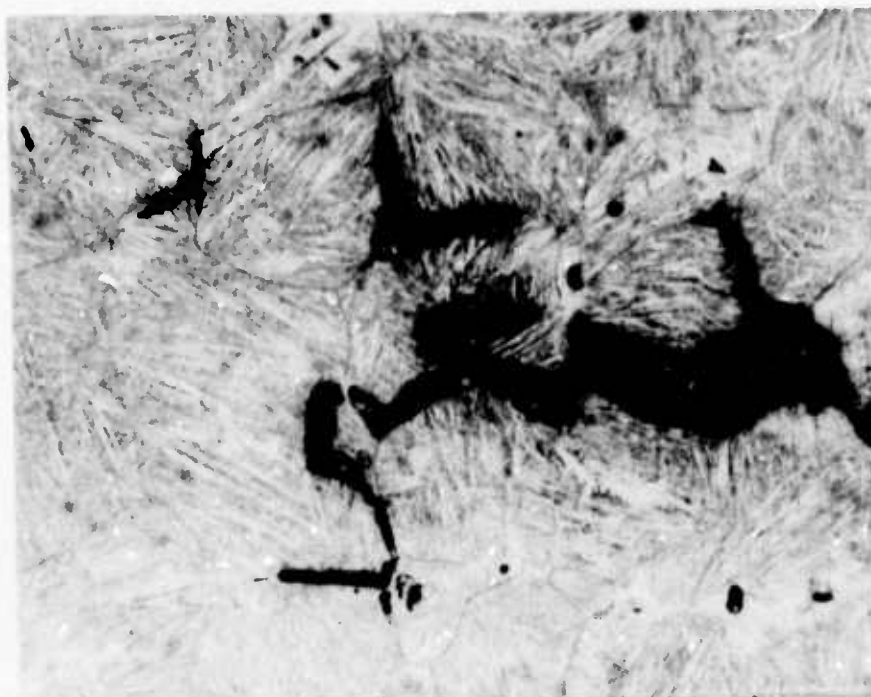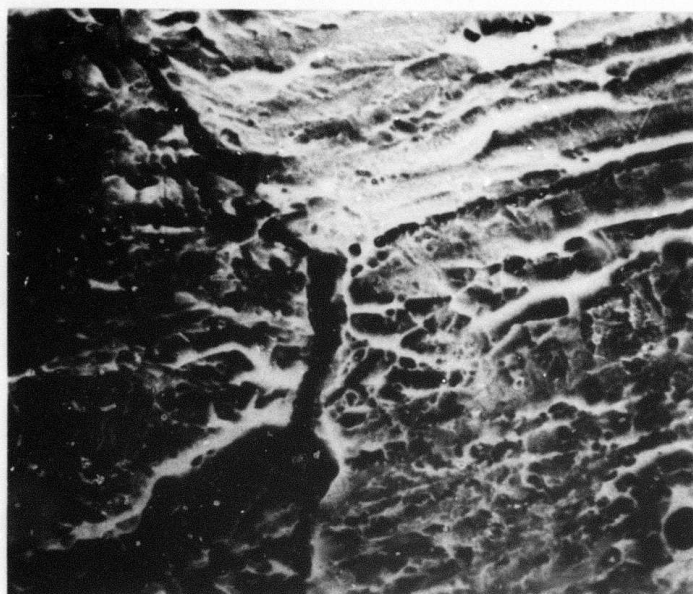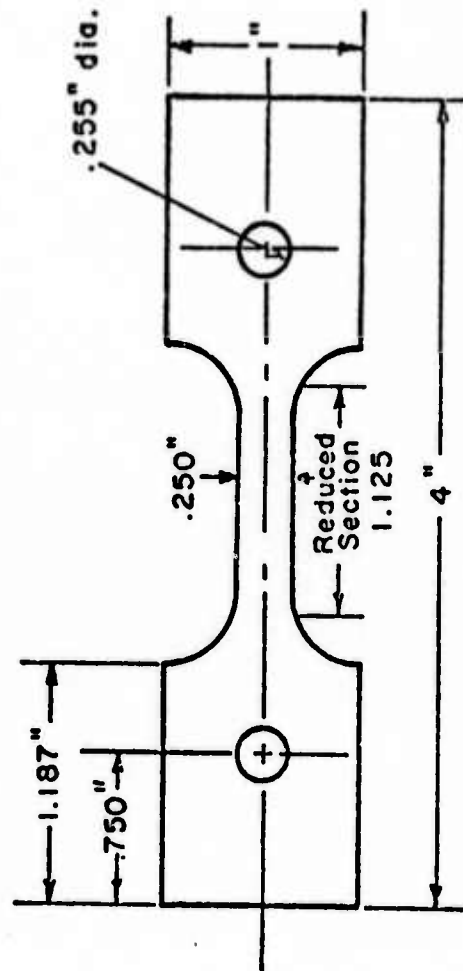Fig. 7B: Fractograph of PS4 Treated as in Fig. 7A.   2000X.

Fig. 8A: Metallographic Section Through Impact Fracture of PS4
Tempered at 750°F (400°C) for 1/2 Hour.

2% Nital Etch   256X

Fig. 8B: Fractograph of PS4 Treated as in Fig. 8A.   2000X.
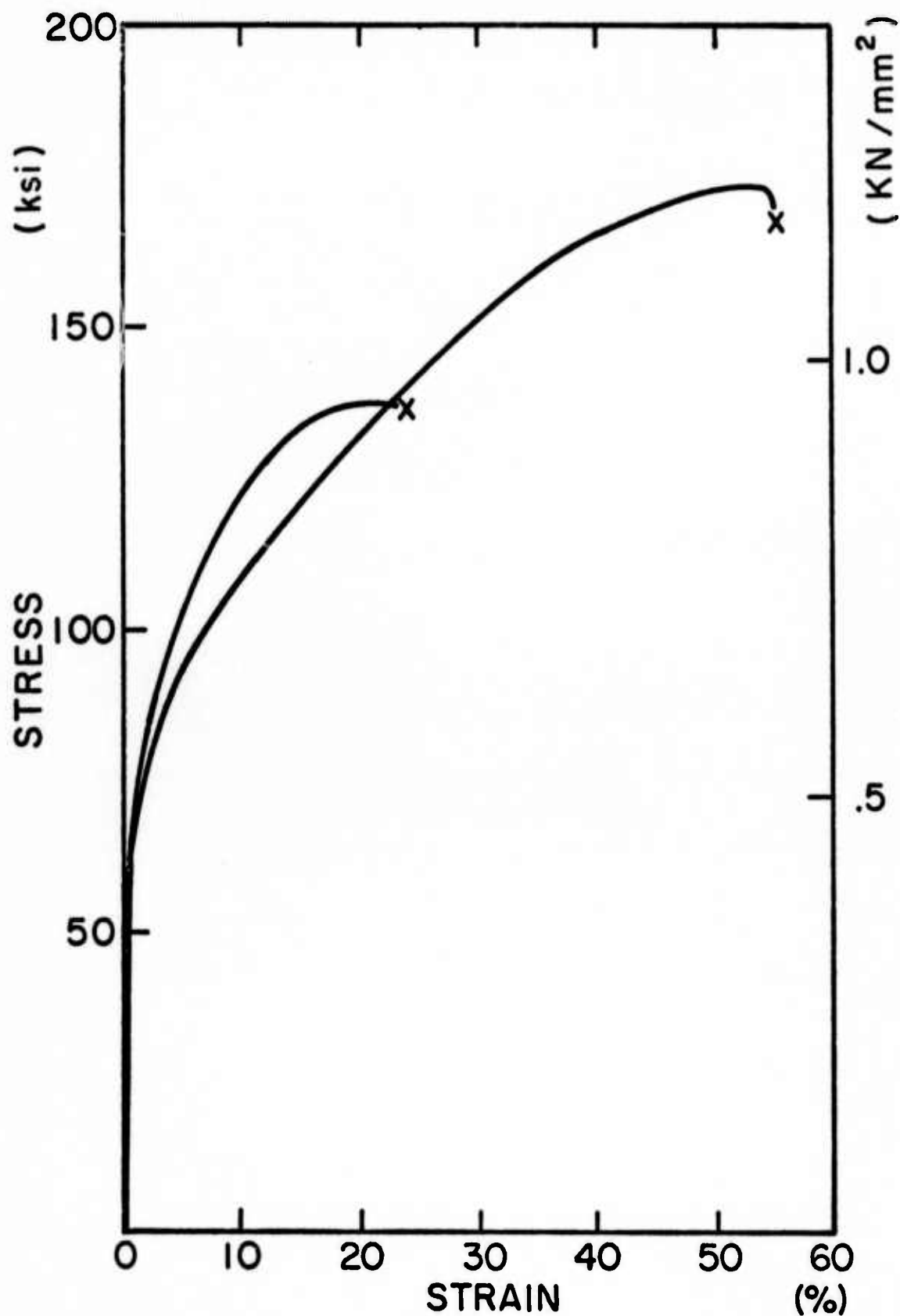
Fig. 9: Tensile Specimen. Nominal thickness 0.080 in. (2 mm).

Fig. 10: Engineering Stress-Strain Curves for Solid Soft Layer Alloy (PS4). Tests were conducted at room temperature at a strain rate of 3×10^{-4} s-1.

low initial yield strength is characteristic of the unstable austenite and their rapid work hardening is accompanied by some serrations on the force-displacement record and by distinctly audible clicks. Following deformation the gauge length of the tensile specimen had transformed to about 50% martensite and near the fracture that amount was significantly higher, up to 80% martensite. The fracture is fairly ductile as shown by the dimples on the scanning electron fractograph, Fig. 11.

V.2. Properties of High Speed Steel (REX 71).

Initially efforts were directed toward maximizing the hardness of the REX 71 and more recently, due to the low effective tensile strength of such material, attention has been redirected to undertempered REX 71 and toward CPM M-4 high speed steel - both of which have reduced hardness but higher toughness. Toward the objective of maximizing hardness, REX 71 samples were given the heat treatment specified by the manufacturer, i.e., austenitized at 2175°F (1190°C), oil quenched and triple tempered at 1000°F (540°C)-- sequence ABCE of Fig. 5. The resulting hardness observed is the manufacturer's published value of R_c 69.6 (1016 DPH). Slight variations in heat treatment resulted in essentially no change in the hardness value. The intercept grain size in REX 71 was about 3.1 μm (corresponding to 41 intersections in 0.005 inches) and also appeared to be independent of small perturbations in heat treatment or rolling procedure. In this condition the fracture path in tension or in impact was almost exclusively along prior austenite grain boundaries. Fig. 12A shows a metallographic section through an impact fracture, and Fig. 12B is a scanning electron micrograph of the corresponding fracture surface. Qualitative observation indicated little energy was required to fracture an unnotched bar of REX 71 in this hard condition.

Tempering at temperatures below that required for maximum hardness is
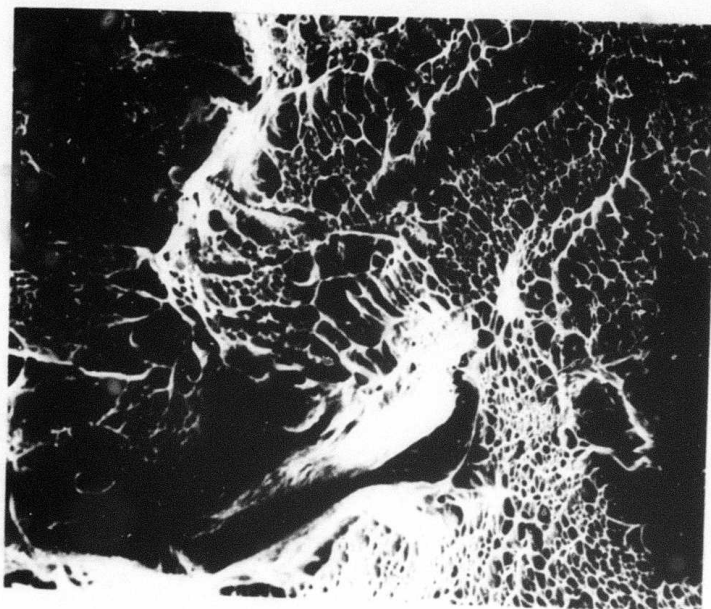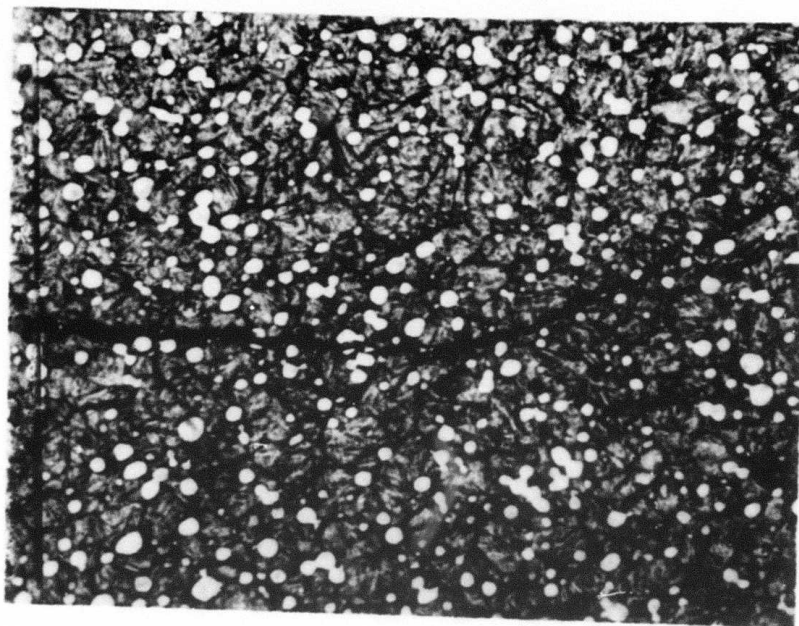
Fig. 11: Tensile Specimen Fractograph SEM PS4.

600X

Fig. 12A: Metallographic Section Through Impact Fracture of REX 71 Triple Tempered at 1000°F (540°C). R_c 69.6

Nital Etch   1280X

Fig. 12B: Fractograph of REX 71 Treated as in Fig. 12A.   2000X.

usually observed to increase toughness more effectively than is overtemper-ing[7]. The hardness of undertempered REX 71 was determined experimentally and is shown in Fig. 13 along with the manufacturer's data on overtempering. As the triple tempering temperature was reduced to 700°F (370°C), impact fracture behavior was qualitatively observed to change to a partially intra-granular path and a qualitative increase in toughness was observed. For the 700°F (370°C) triple tempered sample, Fig. 14A shows a metallographic section through the fracture and Fig. 14B is a scanning electron fractograph. The small change in fracture path shown by comparing Figs. 12 and 14 appears to be accompanied by a toughening.

The increased toughness of the REX 71 developed by undertempering may not suffice to allow this material to be used as a hard layer. To allow for the possibility the properties of M4 steel* are being investigated. The manu-facturer rates[8] the M4 formed from metal powder (CPM) as obtaining a maximum bend strength of 700 ksi in well prepared samples. This compares with scattered bend strengths only about half as large for full hard REX 71 sim-ilarly tested. No essential problems are anticipated in substituting M4 for REX 71 as the hard layer component.

V.3. Properties of Complexes.

The testing of complexes has not progressed as far as had been antici-pated and consequently the properties developed at this stage are not satis-factory. None-the-less some interesting and useful developments have been made and the aim strength and toughness appear to be attainable.

REX 71-PS4 complexes have been prepared with controlled bonding at the

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*A high speed steel whose recommended heat treatment is very similar to that of REX 71. It has 1.35%C, 4.3%Cr, 4%V, 5.0%W and 4.5%Mo which places the carbon activity about 20% lower than that of REX 71 and it is therefore still compatible with PS4.
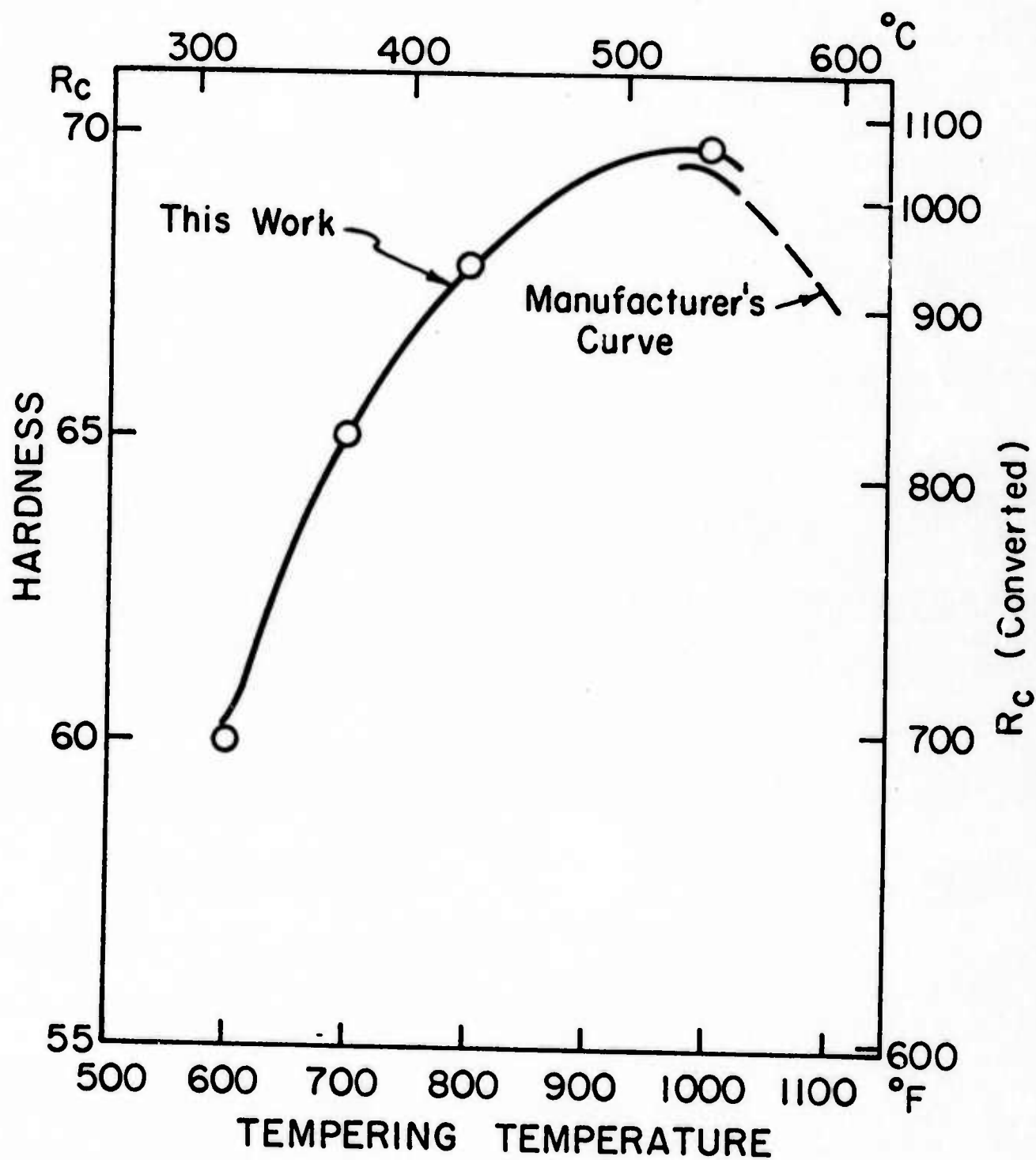
Fig. 13: Hardness of REX 71 Austenitized 2175°F (1190°C) and Triple Tempered for 2 Hours as Shown.
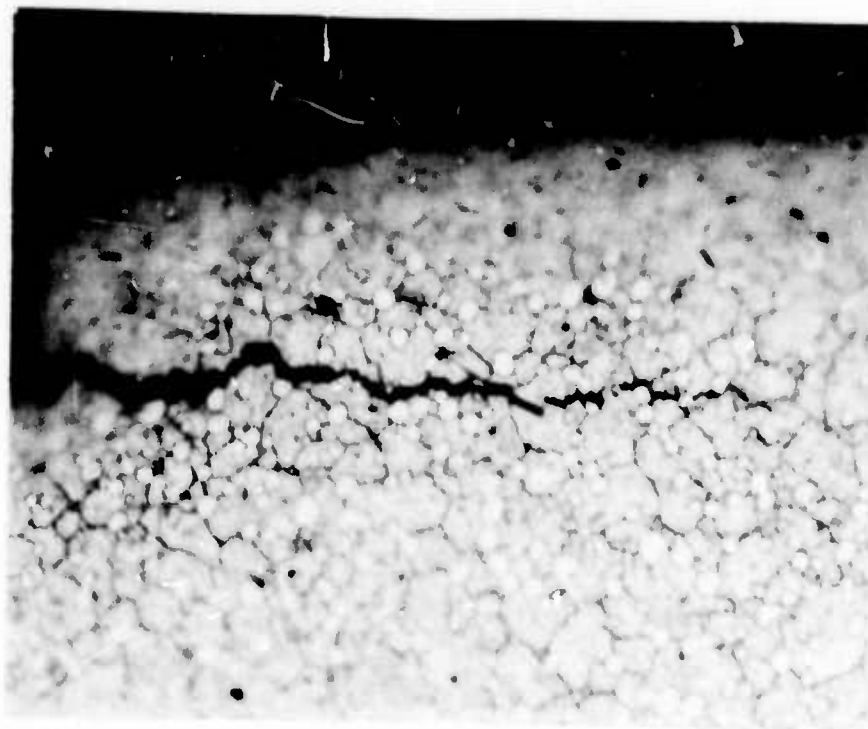
Fig. 14A: Metallographic Section Through Impact Fraction of REX 71
Triple Tempered at 700°F (370°C).

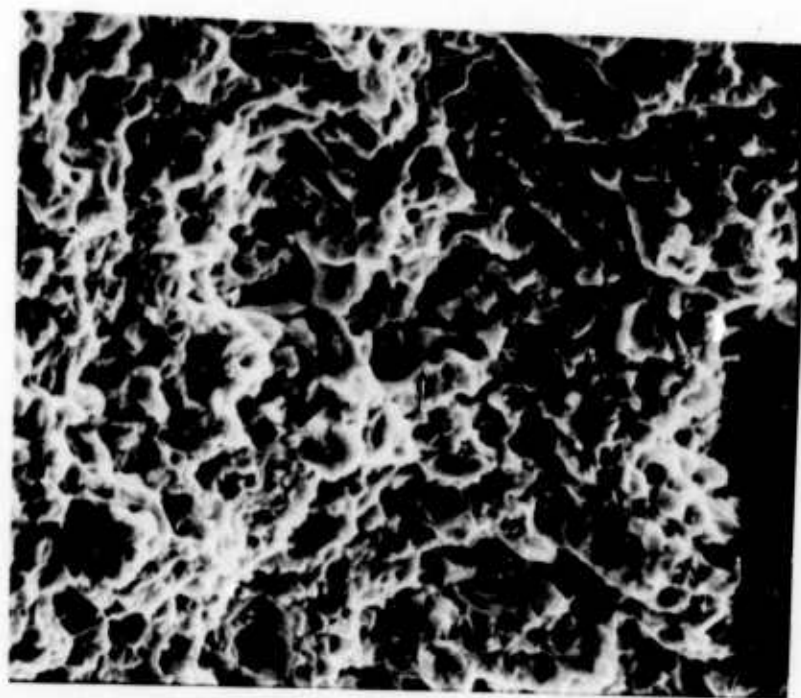Nital Etch   1280X

Fig. 14B: Fractograph of REX 71 Treated as in Fig. 14A.   2000X.

interface and with REX 71 at near maximum hardness. The microhardness pro-
file is shown in Fig. 15. In this figure the horizontal lines are drawn at
the mean hardness levels in all the layers, 900 KHN (at 200 g) for the hard
layers and 200 KHN for the soft layers. The vertical lines represent the
interface zones in this complex. The presence of zones 12 μm thick extending
into the REX 71 layers and the reduction of the maximum hard-layer hardness
from its bulk value of 970 KHN (at 200 g) may be due to decarburization of
the hard layer by the stainless steel used to form the interface. In subse-
quently formed complexes prebonding carburization of the stainless steel foil
to about the carbon potential of the REX 71 has essentially eliminated this
zone. Considering the scatter in the microhardnesses of the hard layers, the
hardness loss from the bulk value is barely significant. It may also be due
to decarburization by the interface foils. The soft layer hardness is essent-
ially unchanged from its bulk value of 222 KHN (200 g) when heat treated
similarly.

For comparison a very recently produced undertempered complex was sec-
tioned and its microhardness profile is shown in Fig. 16. Here the interface
zone is so small (< 8 μm) that it cannot be shown in the profile and the
vertical lines indicate the location of the oxide interface. The hardness of
the soft layer is again essentially the bulk value. In this case the hardness
of the REX 71 layer is the same as the bulk value of 700 KHN (200 g). A
photomicrograph of this section through this complex is shown later including
the fracture profile (Fig. 19).

The tensile behavior of complexes with near maximum hardness in the hard
layers and minimum hardness in the soft layers has been determined at room
temperature. The fracture stress level was low, 163 ksi (1.14 KN/mm^2), in the
two samples tested and only a small amount of plastic strain preceded fracture.
Loading was accompanied by audible clicks in the sample. Most of these

Fig. 15: Hardness Profile of REX 71-PS4 Complex Triple Tempered at 1000°F (540°C) for Maximum Hard Layer Hardness and Minimum Soft Layer Hardness. The soft layer is austenitic because the complex has not been subcooled at any stage in the treatment.
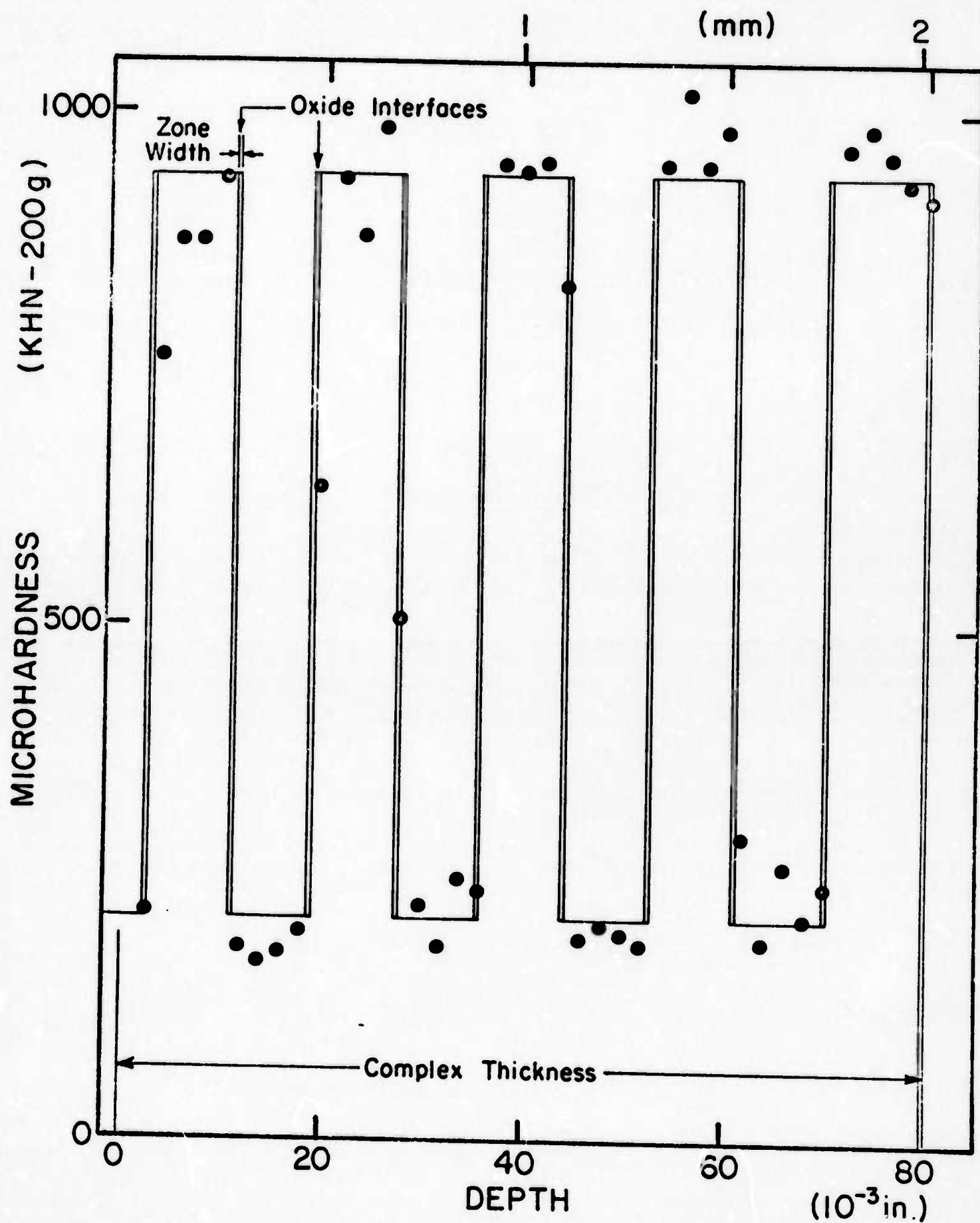
Fig. 15:

Fig. 16: Hardness Profile of REX 71-PS4 Complex Triple Tempered at 700°F (370°C) and not subcooled.
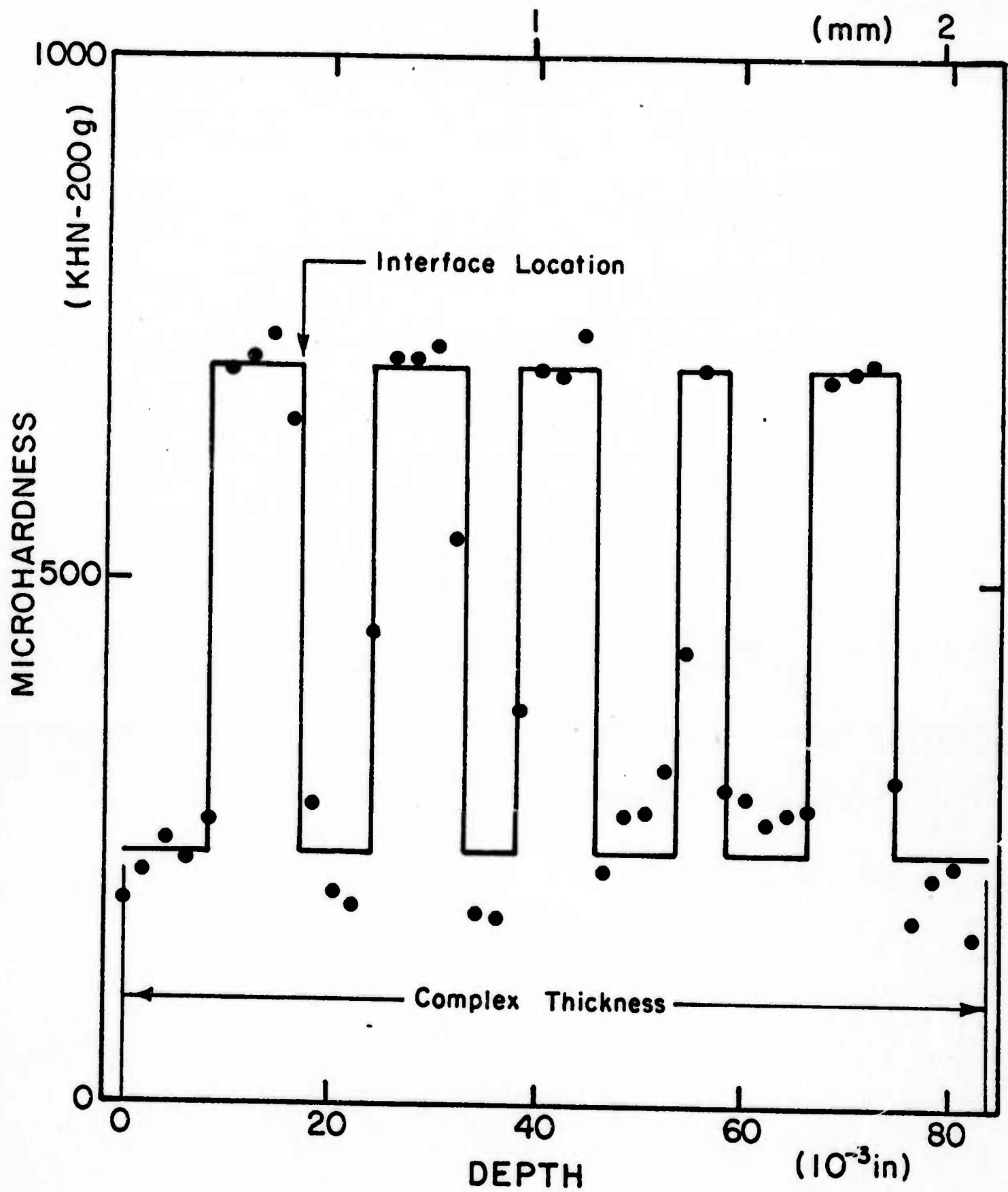
Fig. 16:

apparently were martensite plates forming in the soft layers and some were local cracks in hard layers. As mentioned previously, the hard layers treated to maximum hardness are brittle and fracture along prior austenite grain boundaries. Fig. 17 shows some highly branched cracks in a hard layer. Along most of their length, these cracks follow austenite grain boundaries. These cracks were adjacent to a large crack which crossed one hard layer and was stopped at the interfaces with the adjoining soft layers. Several such cracks were usually present near the ultimate fracture surface. As shown in Fig. 18, these cracks terminate in the interface and become blunted by the combined effect of soft layer deformation and interface cracking. Both these effects can be seen in Fig. 18. Evidently in this complex, parting at interface is sufficiently easy to blunt cracks formed in the hard layers and to inhibit their spreading.

The soft layers transformed to 30% martensite during the straining of the specimen. This transformation was more extensive near the fractures of the hard layers and was especially marked in the vicinity of the complete fracture. The profile of a fracture is illustrated in Fig. 19, which is a section of a tensile fracture of an 700°F (370°C) undertempered complex. The lighter bands are sections through soft, initially austenitic layers. Near the ultimate fracture, parting along the interfaces has allowed necking and appreciable deformation in the soft layers. This deformation indicates that some fracture toughness may be expected. The fracture of the hard layers is still brittle. A scanning electron micrograph of a fracture is shown in Fig. 20. The parting is again evident in this view.

Fig. 17: Section Through a Hard Layer in a Complex Triple Tempered at 1000°F and Not Subcooled. Bonded cracks follow austenite grain boundaries. This section is in the neighborhood of the large crack near the exterior of the lower magnification section in Fig. 18.

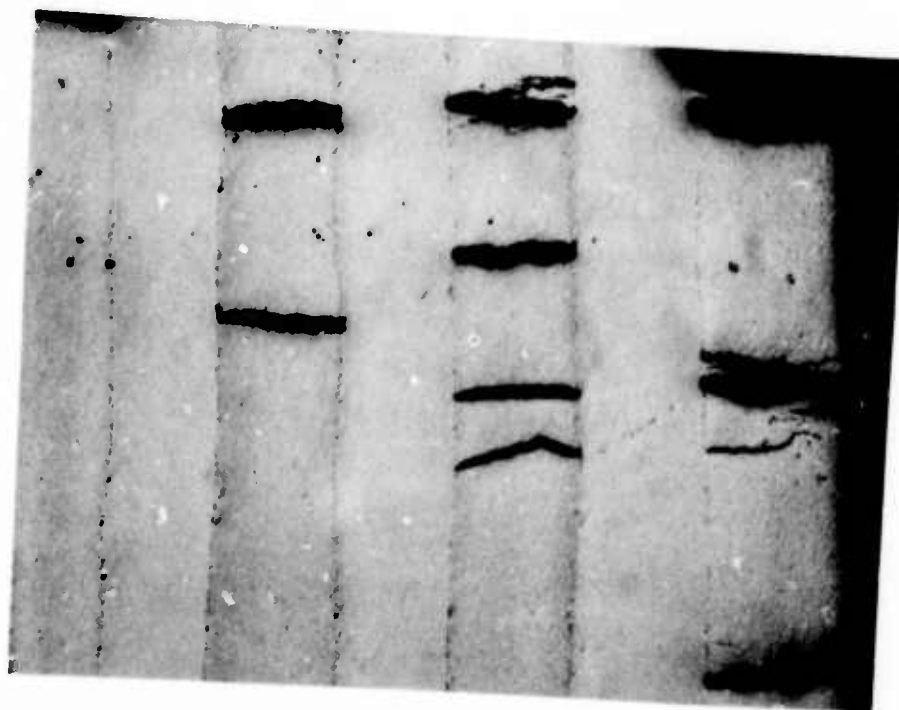Nital Etch   1280X

Fig. 18: Section Through a Hard Layer in a Complex Triple Tempered at
1000°F and Not Subcooled. Notice that many cracks stop at
the oxide interfaces which show up as lightly dotted lines.

Nital Etch   83X
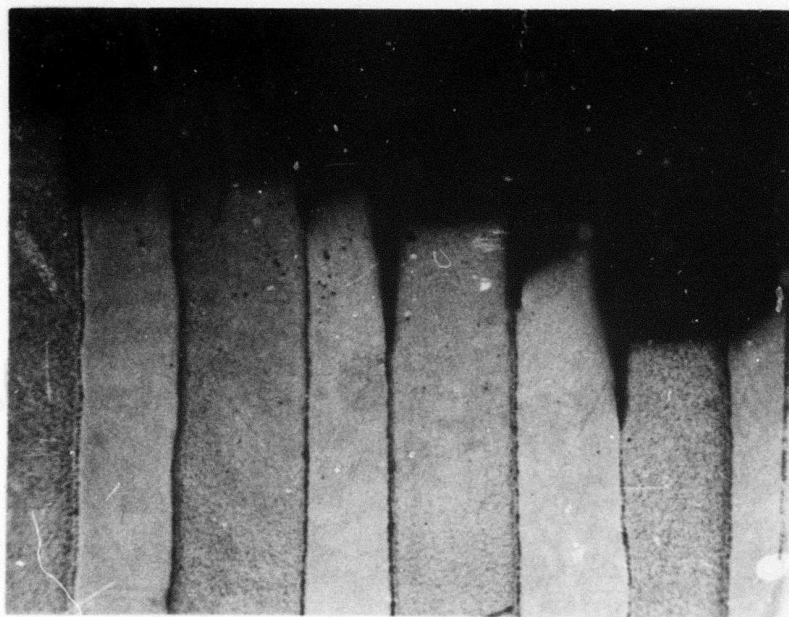
Fig. 19: Profile of a Tensile Fracture of a Complex Triple Tempered
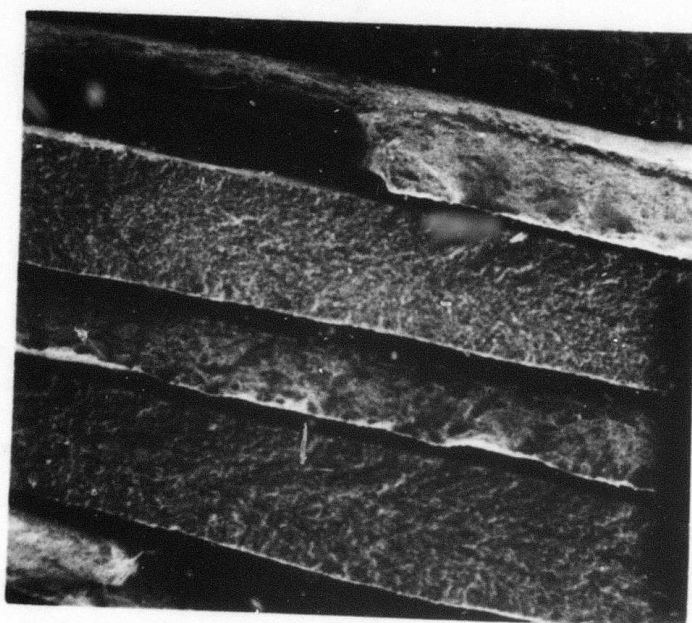at 700°C (370°C). Note splitting at interfaces.

Nital Etch   83X

Fig. 20: Scanning Electron Micrograph of Fracture of Complex
Tensile Specimen Triple Tempered at 700 °F and Not
Subcooled.   120X.

VI. SUMMARY

Complexes of high speed steel and a specially designed compatible soft alloy have been produced and tested in tension. The interfaces between the hard and soft layers are formed from roll-bonded foils of stainless steel carburized and preoxidized on one surface so that the interface developed between the preoxidized surfaces has an energy of 0.02 J/cm^2. Complexes with fully hard REX 71 hard layers do not exhibit good tensile strength because the REX 71 fractures prematurely. Although many cracks in the REX 71 are successfully stopped at the interfaces, sufficient cracking occurs to cause total fracture at a rather low stress (163 ksi: 1.14 KN/mm^2). Tests on undertempered REX 71 and on M4 (fully hardened) are in process. These materials should be less brittle and may (according to bend tests for the M4) have strengths as high as 700 ksi (4.8 KN/mm^2). Ductile fracture in the soft layers appear to have been achieved. This was a requirement for the attainment of fracture toughness in the complex.

Although progress is not as rapid as had been expected, work is continuing toward the design strength-toughness goal of 400 ksi at 150 ksi$\sqrt{in}$, i.e., 2.8 KN/mm^2 at 160 $N/mm^2\sqrt{m}$.

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